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WEAR OF HOMOGENEOUS AND COMPOSITE MATERIALS UNDER CONDITIONS OF--ETC(U)
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United States Air Force

Air Force Office of Scientific Research

AFOSR-76-3087

Wear of Homogeneous and Composite Materials Under Conditions of Repeated Normal and Sliding Impact

Stephen Rice and Roman Solecki

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20. ABSTRACT (Continue on reverse side if necessary and identify by block number)

This report describes experimental and analytical investigations in the area of wear of materials under repetitive compound impact. A reciprocating apparatus which provides controlled, repeatable, measureable impulsive loads between specimen and counterface has been employed in wear testing. Specimen materials in previous experimentation include polymers, composites, and several alloys. That reported herein focusses upon titanium alloys, metallic composites, and copper-chromium alloys. The formation of subsurface zones occurring in impact wear appears to be significant. Correlations between subsurface strain gradients and wear have been over

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demonstrated. These subsurface plastic deformations are dependent upon macroscopic "solid contact" conditions imposed during impact. Temperatures occasioned at and near the wear interface are shown locally to each levels of 800°C for test conditions explored. Void and crack formation in the substrate is occasionally observed, but not nearly to the extent reported for materials under conditions of sliding wear. Analytical investigations are directed toward determination of the states of stress in material specimens undergoing controlled repetitive impact. Results are derived from the equations of the linear theory of elasticity applied to a finite cylinder impacting a rough, rigid surface. Papkovich-Neuber functions are used to solve the static problem and Lame potentials - to solve the dynamic problem. Correlations between subsurface stress and wear debris formation is sought.

Abstract

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Analytical investigations are directed toward determination of the states of stress in material specimens undergoing controlled repetitive impact. Results are derived from the equations of the linear theory of elasticity applied to a finite cylinder impacting a rough, rigid surface. Papkovich-Neuber functions are used to solve the static problem and Lamé potentials - to solve the dynamic problem. Correlation between subsurface stress and wear debris formation is sought.

Experimental

As noted previously in Interim Reports, a wide range of materials of technological significance has been studied [1,2]. In this work, experimental investigations are directed toward elucidating processes or mechanisms by which materials wear under conditions of repetitive impulsive loading. The research has thus far confirmed the significance of gross external variables (peak stress levels, relative sliding velocities), and it has demonstrated that microstructural and chemomechanical factors are at least as important as macroscopic material properties such as hardness (within the range of test parameters explored).

The format to be followed herein is that salient experimental results will be summarized by specimen material category. It should be noted that testing has been standardized so that, to date, all specimens have been repetitively impacted against a 17-4 PH steel counterface. Following this explication by material category, commonalities and important differentiating factors will be noted. This then leads to a statement of current states of understanding and directions for extended near-term experimental investigations.

A. Titanium Alloys

Three alloys have been tested against the standard 17-4 PH stainless steel counterface. These are: (1) IMI-685, an alloy having good creep resistance at high temperature, and consisting of a predominantly alpha phase microstructure; (2) RMI-5522S, an alloy tested in a condition of high strength and good ductility but poor creep resistance, and also consisting of a predominantly alpha phase microstructure, but differing considerably from the IMI 685 in morphology; and (3) RMI-beta, a non-commercial alloy consisting

of essentially all beta phase. Composition, heat treatment prior to wear testing, and supplier data are presented in [3].

For each alloy tested, the relative transverse sliding velocity was varied at least over the range 2.1 to 6.3 m/s and weight loss measurements performed. Some of these data are shown in Figure 1, which summarizes results from tests in which the nominal peak normal impulsive stress and test duration are fixed at 18.6 MPa and 150,000 repeated compound impulsive load cycles respectively. From the figure it is apparent that the two predominantly alpha phase alloys behave similarly, but the beta alloy gives rise to significantly different wear behavior. Further, this difference is not in the wear of the beta alloy (titanium test specimen), but rather in the coun rface (17-4 PH steel). This demonstrates the importance of monitoring system changes in experimental wear studies.

The second significant feature depicted in Figure 1 is a marked velocity dependence. In this, all three alloys appear to exhibit the same behavior, with the beta alloy giving rise to more extreme variations. In short, there appears to exist a minimum in the velocity range near 4.2 m/s. To determine whether or not such a minimum exists for other test durations, additional experiments were conducted. Figure 2 depicts results obtained for totals of 50,000 and 1,000,000 repeated load cycles. These data show a clear maximum at the 2.1 m/s velocity for both specimen and counterface.

The above results led to further analysis and testing. Figure 3 presents results from an experiment designed to explore the feasibility of in situ surface improvement by virtue of repetitive contact under specified conditions. Detailed methodology is presented in [3]; here,

it suffices to note that the objective was to conduct a test at high relative sliding velocity (10.5 m/s) to attempt to "stabilize" the near surface material zones which form during repetitive high speed contact. Such potential for stabilization can be inferred from Figure 2 in which less wear was observed at high sliding velocity. Unfortunately, the "stabilization" process appears to have been unsuccessful for the conditions explored, in that when a specimen/counterface pair which had been "run-in" at 10.5 m/s was subsequently tested at 2.1 m/s, the characteristically higher wear rate of the lower velocity was again observed.

To analyze specific phenomena (such as the velocity dependencies noted above), a variety of procedures are utilized. These include surface and subsurface examination via light and scanning electron microscopy, as well as element mapping by means of energy dispersive X-ray techniques. A typical EDX result for the RMI beta alloy is shown in Figure 4 and the data are suggestive of processes giving rise to velocity dependent wear rates. The general observation is that a varying degree of material is transferred from specimen to counterface, and vice versa. Indeed, both the extent and direction of material transport depend upon the magnitude of the relative transverse sliding velocity occurring during the impact event. Such results can be quantified for purposes of explication. A typical set of curves is shown in Figure 5.

Representative micrographs are included as Figures 6 - 11. Figures 6 and 7 show subsurface sections of worn RMI 5522S specimens. Both figures indicate the characteristic formation of subsurface zones as documented in earlier research reports [2,4]. It is clear that near-surface zones of fundamentally different microstructure are produced in situ, with Figure 6 revealing that such layer formation can occur following as few

as 250 repeated load cycles. As Figure 7 indicates, both near-surface layer morphology and the extent of subsurface deformation are altered with increasing number of load cycles.

Figures 8 and 9 are indicative of surface features following wear of the pair RMI 5522S titanium alloy specimen against 17-4 PH stainless steel counterface. Figure 8 shows the specimen; plowing grooves and lamellar wear sheets are clearly defined. Figure 9 shows the counterface, and the deposit of transferred material (with a characteristically layered morphology) over a previously plowed area. Note both Figures 8 and 9 illustrate worn surface appearance from a material pair tested at 2.1 m/s. The worn surfaces produced from experiments conducted at 4.2 m/s are decidedly different, as can be seen in Figures 10 and 11, showing specimen and counterface respectively. The difference between the worn surfaces produced at the two velocity levels underlines the significance of the combined effects of gross external variables (stress, velocity) and chemo-mechanical material processing which occurs in situ. From observation of oxide film formation with variation in color from bright blue to straw yellow, surface temperatures during testing are estimated to have reached roughly 800°C.

Further results on work with titanium alloys are reported in [5]. In this, one of the primary findings is that microstructural refinement which may not be resolvable by scanning electron microscopy is observable by means of transmission electron microscopy utilizing replicas from subsurface sections. A second significant finding is that near-surface material zones produced in repetitive impact wear remain in the crystalline state, as determined by TEM studies of near surface foils. This finding is in agreement with the wo

of others [6,7] who have studied near surface material zones produced under conditions of sliding (no impact) wear.

B. In Situ Composites

Testing has been conducted on several metallic composites produced by controlled solidification of eutectic alloys. These materials are characterized by aligned interpenetrating phases, and possess excellent strength at elevated temperatures and good resistance to creep.

At present, an experimental investigation on cobalt base directional composites has been completed [8] and results are outlined herein. Two such materials, both featuring rather brittle TaC fibers contained within comparatively ductile support matrices (Co-Ni or Co-Ni-Cr), have been tested against the standard 17-4 PH steel counterface.

As is well known, fiber orientation is an important parameter not only in terms of strength but also in terms of wear [9,10,11]. Accordingly, two fiber orientation modes were utilized in impact wear testing, as described in Figure 12. Some weight loss data are presented in Figure 13 where results have been obtained for test durations of 150,000 repetitive load cycles at a fixed level of nominal peak normal impulsive stress (18.6 MPa). The velocity dependence is evident from these data, but the effect of fiber orientation is seen to be negligible.

Figure 14, however, depicts results for test durations beyond 150,000 load cycles. While counterface (17-4 PH steel) wear is insensitive to extended load cycling, the fiber orientation effect upon specimen wear is clearly significant. The reason for the increased wear resistance of the Mode 1 configuration is probably identical to that proposed for impact wear of graphite epoxy composites [9].

As with the titanium alloys, microscopy and X-ray studies were performed with the metallic composites. Figure 15 summarizes some energy dispersive X-ray analysis results to allow interpretation of material transport. These results can then be utilized in concert with surface and subsurface (sections through specimen and counterface) microscopy to allow insights into near-surface layer formation, and ultimately to wear particle formation mechanisms.

Figure 16, for example, shows a subsurface section from a Cotac specimen. One notes the three characteristic zones discussed earlier [2]; i.e., zone 1 = undisturbed substrate; zone 2 = observably plastically deformed intermediate "layer"; and zone 3 = morphologically and crystallographically distinct near-surface layer. With the metallic composite, the brittle TaC fibers can be seen to have fractured into progressively finer lengths, and to have become increasingly more aligned with the sliding direction as the wear interface is approached.

There is a strong suggestion of a crack or cracks tending to separate Zone 3 from Zone 2. Thus, a wear sheet of approximately 5 microns thickness could be in a state of impending delamination. It is fundamentally significant, however, that Zone 3 and Zone 2 are <u>different</u> materials, with Zone 3 having been produced in situ.

C. Copper Alloys

Following the work of Pamies-Teixeira, Saka and Suh [12], a series of tests was conducted to assess the effect of solute atoms on impact wear. Two copper-chromium alloys were supplied from this research group, and were investigated under conditions of compound impact wear. The atomic

% Cr in the two alloys is roughly 0.58 and 0.81, with the former concentration giving rise to less substitutional solid solutionizing with consequent lesser hardness. It should be noted, however, that the measurable effect of atomic concentration on hardness for these two alloys is extremely small.

Testing was conducted at a variety of relative transverse sliding velocities and nominal peak normal impulsive stress levels, viz. V = 0.5 and 2.1 m/s; σ = 3.7, 7.2 and 11.2 MPa. In general, wear increased with increasing stress, and more wear occurred at the lower test velocity than at the higher for all stress levels. Test duration was typically 250,000 repetitive load cycles, with weight loss data points taken at intervals of 50,000 cycles. No significant differences were observed between the two Cu-Cr alloys.

A priori it had been intended to utilize quantitative metallographic procedures to describe void and crack content following impact wear. However, no features of this type could be resolved with either the light or scanning electron microscope. Nonetheless, there was considerable evidence of subsurface plastic deformation, which did permit quantitative assessment of this feature which is, for impact wear under the test conditions explored, a far more pervasive characteristic than is the presence of subsurface voids and cracks. Measurement of plastic deformation was accomplished by noting displacements of subsurface particles relative to the unaffected (Zone 1) base microstructure. This involved a procedure whereby SEM micrographs were taken at the same magnification over the same relative area on each specimen. These micrographs were enlarged to 10x13" prints over which a grid was placed to facilitate measurement. Data points were

established from the wear surface to a depth of approximately 60 microns in 8-10 micron steps. Since measurements of this type are necessarily somewhat subjective, a minimum of three areas on each photo was taken and points were co-plotted. From these data, the "best fit" curve was established. All such measurements were made from specimen cross-sections at the geometrical center of the section. Thus, leading and trailing edge effects are eliminated [in general, the depth of plastic deformation is minimum at the leading edge and increases rather linearly toward the trailing edge, where contact ceases and often a severely plastically deformed "wear lip" is observed].

Detailed elaboration of procedures and results is presented in [13]. For the present, it suffices to note the following. Figure 17 is representative of surface morphology, and is typical of results obtained with other metallic materials. Figure 18 shows representative subsurface features; as noted previously, *three (at most) characteristic zones* can be identified. Figure 19, obtained following meticulous polishing and etching procedures, reveals some of the microstructural features of Zone 3, which are normally difficult to resolve with light or scanning electron microscopy. The continuum of grain refinement from Zone 2 into Zone 3 is evident. No void or crack formation is observed; if these exist, they are below the resolution of the SEM, viz. $\sim$ 200Å.

Figures 20 and 21 are illustrative of localized layering phenomena which occur as a result of material transport (transfer) during in situ formation of near-surface material structures (Zones). Figure 20 is a subsurface SEM micrograph, and Figure 21 is the corresponding EDX element map of chromium rich areas. In this case, the specimen is Cu 0.58 Cr and the standard counterface is 17-4 PH stainless steel (Fe-17Cr-4Ni-4Cu). The steel is richer in chromium than is the copper alloy, and Figures 20 and 21

indicate not only the surface layer (on the Cu-Cr specimen) but also a subsurface layer (in the Cu-Cr specimen) to consist of predominantly counterface material. This result indicates the significance of in situ material transfer in modifying materials in the wear interface and into the near surface zones. In addition, it should be noted that all qualitative observations made for material transport in wear of the Cu-Cr alloys correspond directly to the quantitative analyses performed for Ti alloys.

Figures 22 and 23 show preliminary results on subsurface deformation gradients for the Cu 0.58 Cr alloy following 250,000 repetitive load cycles. Both velocity and stress dependencies are clear, and correspond to results noted earlier based upon weight loss measurements. In sum, at lower relative transverse sliding velocity, there is greater plastic deformation at a given depth from the surface than at higher velocity. The reason for this is thought to be due to lower frictional tractive force transmission in thermally softened zones, with higher temperatures having been occasioned at the higher relative sliding velocities.

The second finding of significance as evidenced in Figures 22 and 23 is that as stress levels are increased, there is greater plastic deformation at a given depth from the surface than at lower stress levels. This result is straightforwardly rationalized in terms of larger frictional tractive forces being occasioned with higher levels of imposed nominal impulsive stress.

Summary of Experimental Results

As noted previously [2], the formation of subsurface zones occurring in impact wear appears to be significant. The present work has demonstrated correlations between subsurface strain gradients and wear. Further, these subsurface strain gradients are seen to be dependent upon the macroscopically imposed conditions of the "solid contact" occurring during impact.

Temperatures occasioned at and near the wear interface are known to be high (at least of the order of 800°C). Laboratory attempts to accurately measure these surface temperatures, in situ under conditions of repetitive impulsive contact, are continuing. This is a difficult task. These surface temperatures are clearly of considerable importance in establishing near surface material zones, and are obviously a strong function of imposed relative sliding velocity. Thus qualitatively it is clear that both stress levels and velocity are important "external" variables in impact wear processes.

Subsurface void and crack formation of the type observed by Suh and others for sliding wear is occasionally observed in subsurface sections of specimens having been subject to impact wear. If a delamination process is occurring in impact wear, it is serving to separate Zone 3 material from Zone 2. Therefore, a complete description of the impact wear process must include an elaboration of mechanisms of formation of near-surface material layers.

Analytical

During the third phase of our work the effort was directed toward three goals:

(1) completion of the analysis of the controlled impact of a one-dimensional, homogeneous, finite elastic rod upon an elastic half-space,

(2) continuation of the analysis of the static model of the controlled impact of a three-dimensional elastic cylinder on a moving, rough, rigid surface,

(3) initiation of the analysis of the impact of a three-dimensional elastic cylinder (propagation of stress waves taken into account) on a moving, rough, rigid surface.

As was described elsewhere [2] analysis of the rod was a preliminary work designed to investigate procedures useful when considering the more realistic three-dimensional description of the problem of controlled impact, as well as to acquire a better understanding of the mechanical behavior of a system under conditions of controlled impact.

The results of this investigation have been presented (see abstract [14]) and published [15].

As reported earlier [2] it was found that the axial stress at any point of the rod results from three components: 1) a stress wave proportional to the velocity of the impact and composed to two rectangular waves propagating from the striking end and from the far end; 2) a stress wave generated by the signal f(t), i.e., the controlled displacement of the far end; 3) a stress wave resulting from the elastic flexibility of the support.

For the final version of this analysis the original attempt of inverting the Laplace transform of the unknown displacement was abandoned. Instead, in order to simplify the procedure, Eason's response function [16] was approximated by a bilinear function (as explained in [15]). This time the inversion

of the Laplace transform did not present unmanageable difficulties, and the numerical results have confirmed the intuitive expectation that the flexibility of the half-space has a moderating influence on both the displacements and the stresses.

(2) In this part of the project it is attempted to find the effect of friction on the stress distribution of an elastic, circular cylinder pressed against a moving, single, rigid surface [17]. This is a quasi-static approximation of the dynamic problem of the controlled impact of the cylinder on a moving, rough, rigid surface. The presence at the lower end of shear stresses due to friction destroys the axial symmetry of the problem, thus introducing considerable additional difficulties. Partial uncoupling of Navier's equations in the polar coordinate system was achieved by using the Papkovich-Neuber functions. Consequent application of double finite Fourier transformations with respect to the angular and axial variables resulted in complete uncoupling of the transformed differential equations. Application of the boundary conditions at the stress-free lateral surface of the cylinder and of the conditions at its flat ends leads, after numerous, extremely cumbersome rearrangements to a system of coupled integral equations with respect to the transforms of the shear stresses at the flat ends. This system is being solved by the method of collocation. However, since the upper surface of the cylinder is prevented from undergoing in-plane displacements and the adjacent lateral surface is stress-free, a singularity results at the edge of the upper surface [18]. This causes additional difficulties which are overcome by dividing each of the unknown discontinuous functions into two parts: a) a regular part, b) a singular part in which the strength of the singularity is known but the intensity is unknown. This part of the work is nearing completion.

(3) The analysis of the controlled impact of the cylinder on the moving rough, rigid surface is analogous to the analysis of quasi-static problem. Obviously, though, the time-dependence makes the problem more difficult. The Laplace transform was used in order to eliminate the time variable. Uncoupling was achieved, in the dynamical problem, by means of the Lamé potentials. The double finite Fourier transform was employed to eliminate the dependence on the angular and axial variables. The application of the available boundary conditions will result in a coupled system of integral equations. Their solution will pose the additional difficulty of inverting the Laplace transformation. It is hoped that the procedure recently suggested by Talbot [19] will be of help in this matter.

Acknowledgement

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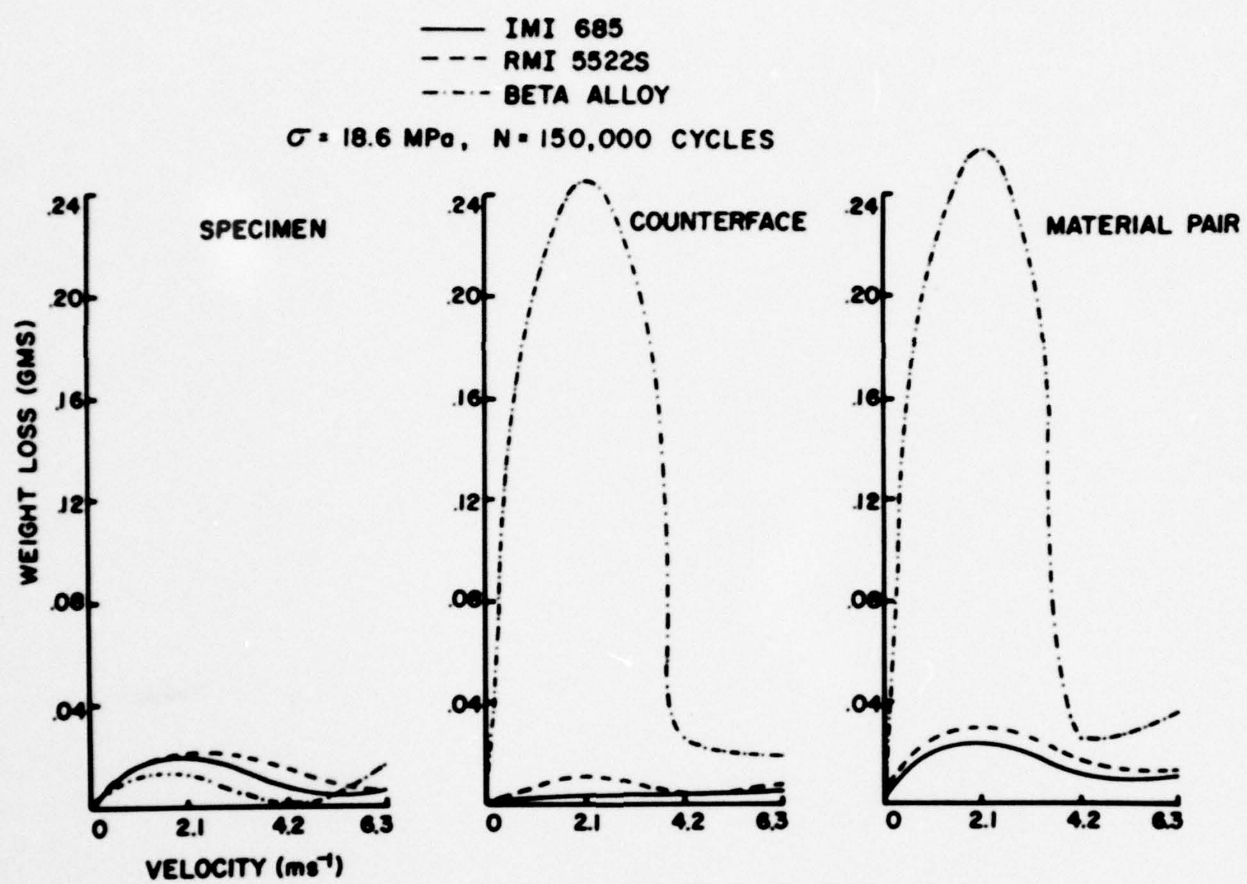

IMI 685
RMI 5522S
BETA ALLOY
σ = 18.6 MPa, N = 150,000 CYCLES
SPECIMEN
COUNTERFACE
MATERIAL PAIR
WEIGHT LOSS (GMS)
.24
.20
.16
.12
.08
.04
0
2.1
4.2
6.3
VELOCITY (ms⁻¹)

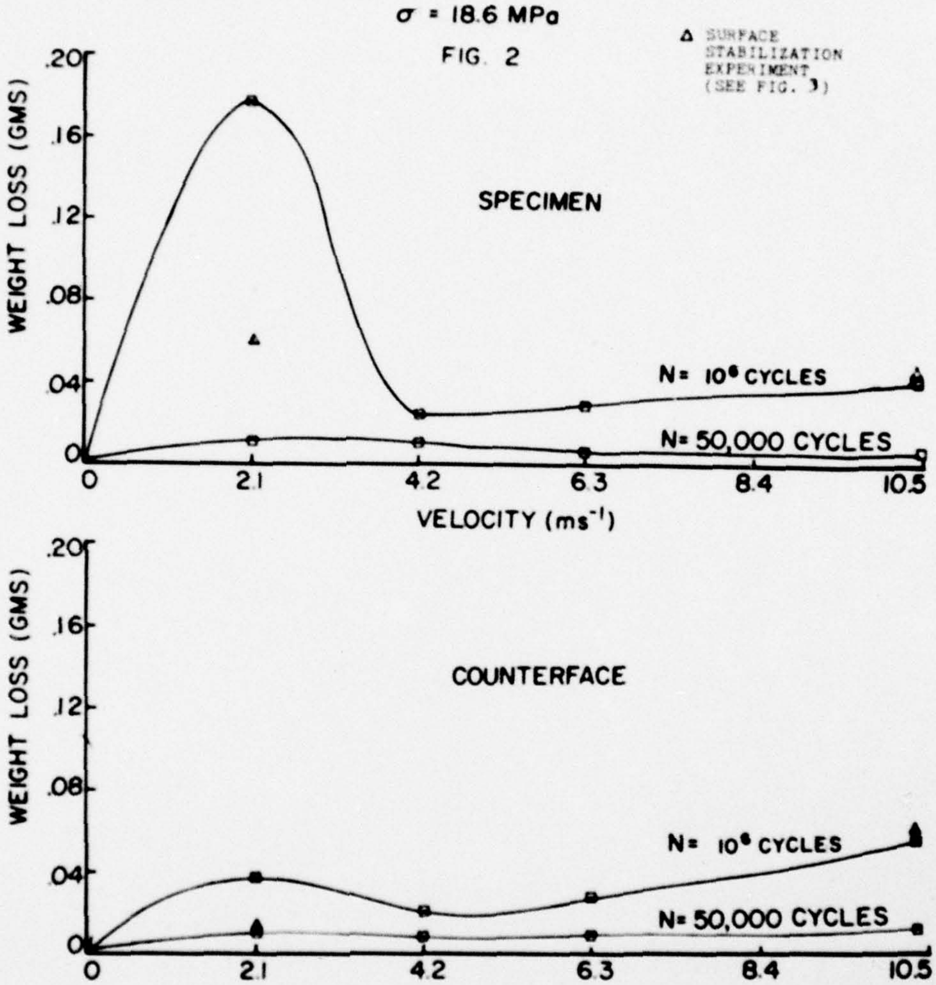
FIGURE 1

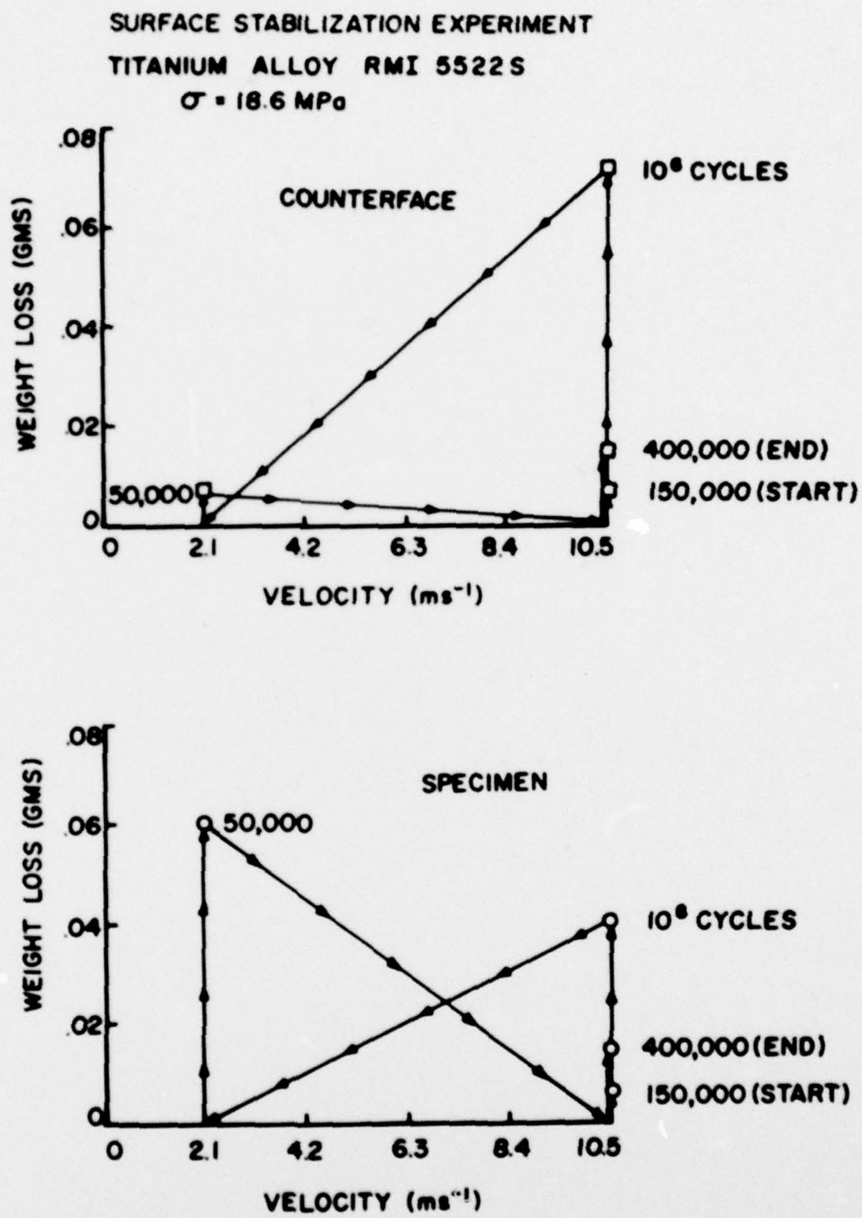
EXTENDED CYCLE TESTING

TITANIUM ALLOY RMI 5522S

σ = 18.6 MPa

FIG. 2



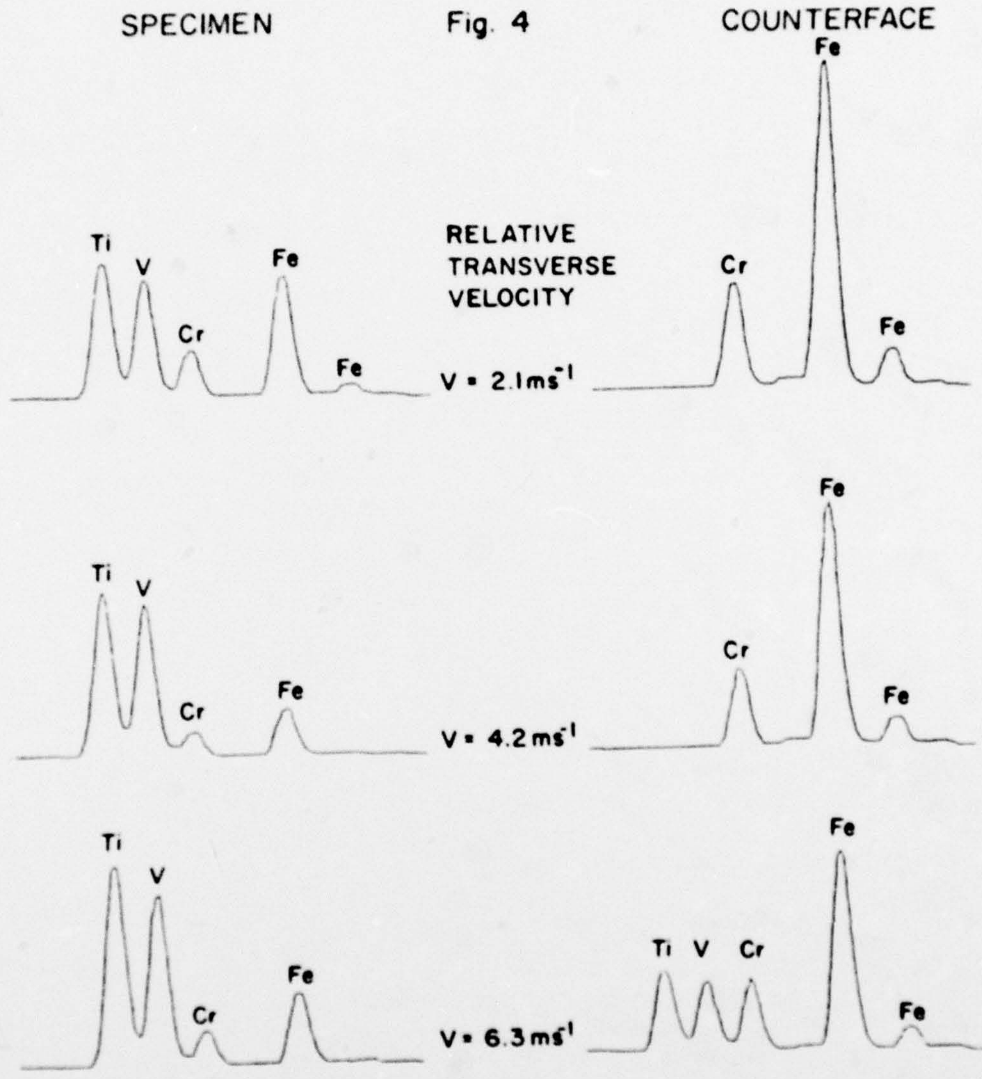


SURFACE STABILIZATION EXPERIMENT
TITANIUM ALLOY RMI 5522S
σ = 18.6 MPa
COUNTERFACE
WEIGHT LOSS (GMS)
.08
.06
.04
.02
0
0
2.1
4.2
6.3
8.4
10.5
VELOCITY (ms⁻¹)
10⁶ CYCLES
400,000 (END)
150,000 (START)
50,000
SPECIMEN
WEIGHT LOSS (GMS)
50,000
10⁶ CYCLES
400,000 (END)
150,000 (START)
VELOCITY (ms⁻¹)

Fig. 3

ENERGY DISPERSIVE X-RAY RESULTS

RMI BETA TITANIUM



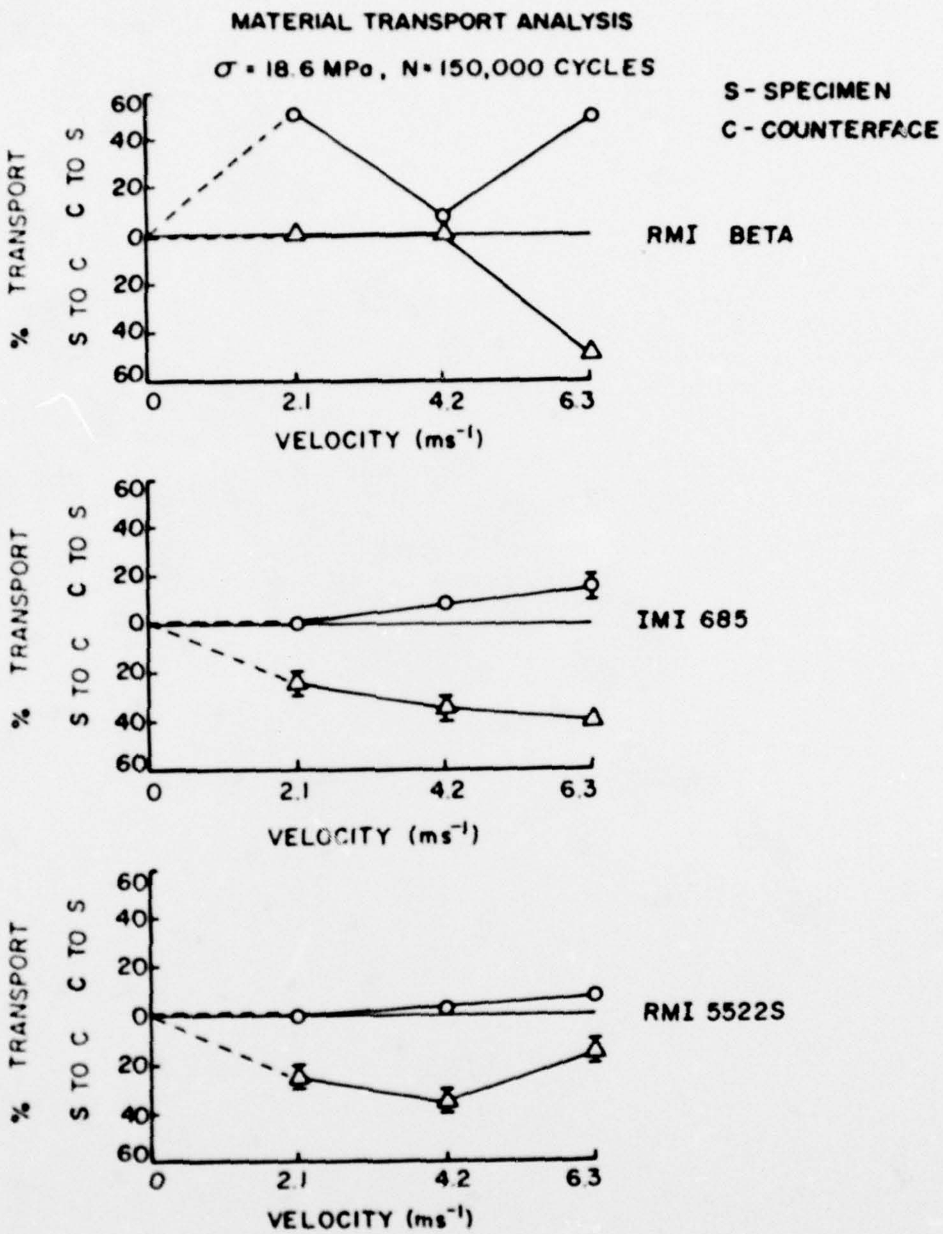

MATERIAL TRANSPORT ANALYSIS
σ = 18.6 MPa, N = 150,000 CYCLES
S - SPECIMEN
C - COUNTERFACE
% TRANSPORT
S TO C
C TO S
RMI BETA
IMI 685
RMI 5522S
VELOCITY (ms^{-1})
0
2.1
4.2
6.3
60
40
20

FIG. 5

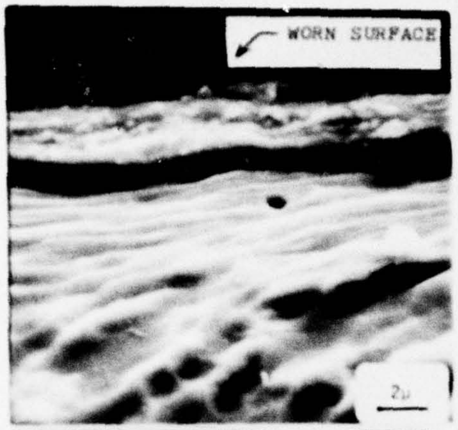


FIGURE 6 Micrograph (SEM) of RMI 5522S, worn surface cross section, note the formation of layers in the near surface zone. ($V=2.1$ ms^{-1}, $\sigma=18.6$ MPa, $N=250$ cycles).

FIGURE 7 Micrograph (SEM) of RMI 5522S, worn surface cross section, note the break up of surface layers. ($V=4.2$ ms^{-1}, $\sigma=18.6$ MPa, $N=10^6$ cycles).

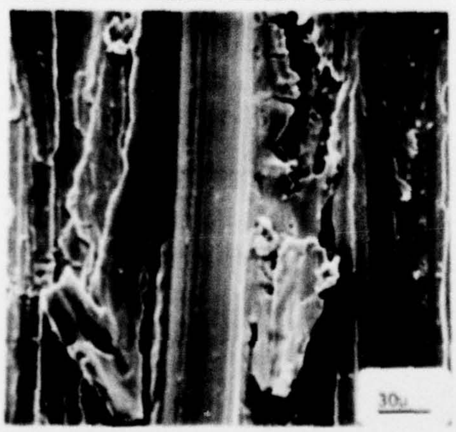


FIGURE 8 Micrograph (SEM) of RMI 5522S, worn surface. ($V=2.1$ ms^{-1}, $\sigma=18.6$ MPa $N=10^6$ cycles).

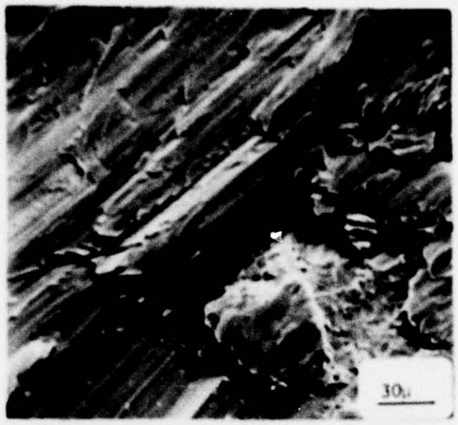


FIGURE 9 Micrograph (SEM) of 17-4 PH counterface, worn surface. (V=2.1 ms^{-1}, σ=18.6 MPa, N=10^6cycles).

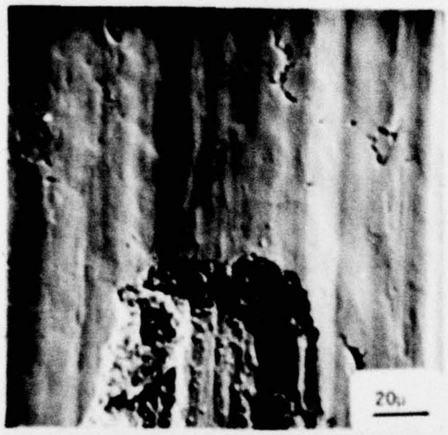


FIGURE 10 Micrograph (SEM) of RMI 5522S, worn surface. (V=4.2 ms^{-1}, σ=18.6 MPa, N=10^6 cycles).

FIGURE 11 Micrograph (SEM) of 17-4 PH counterface, worn surface. (V=4.2 ms^{-1}, σ=18.6 MPa, N=10^6 cycles).

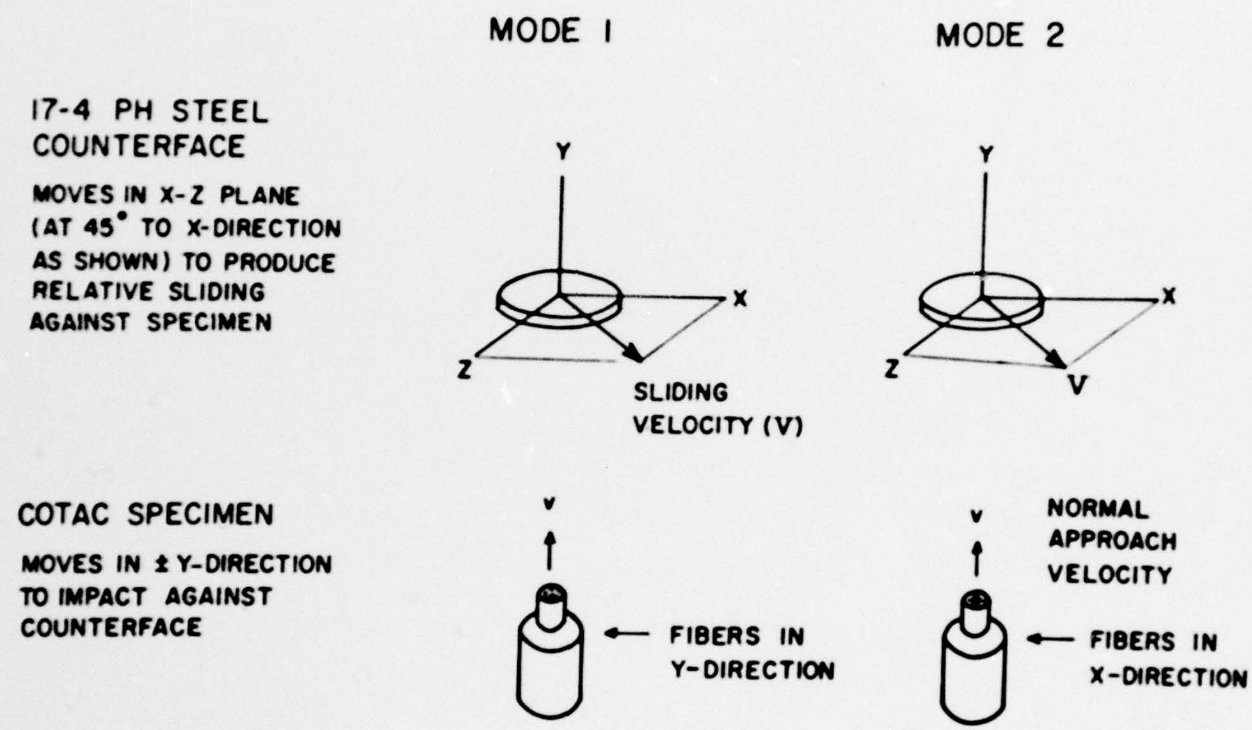
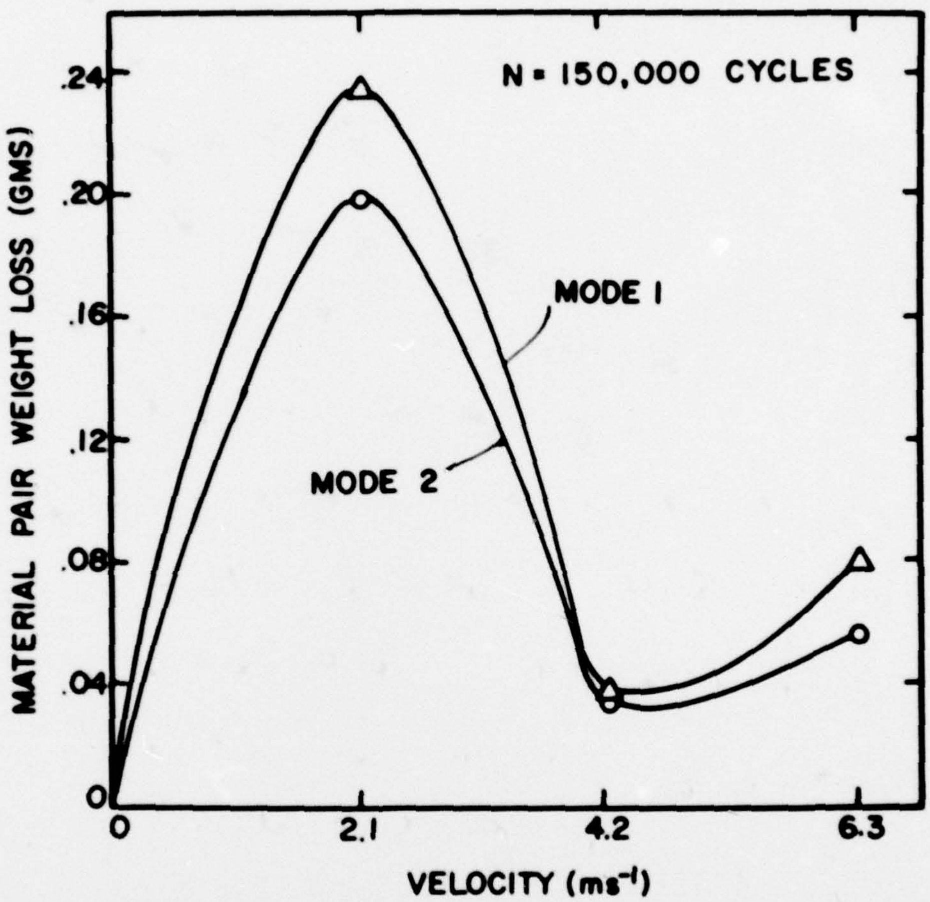

MODE 1
MODE 2
17-4 PH STEEL COUNTERFACE
MOVES IN X-Z PLANE (AT 45° TO X-DIRECTION AS SHOWN) TO PRODUCE RELATIVE SLIDING AGAINST SPECIMEN
Y
X
Z
SLIDING VELOCITY (V)
V
COTAC SPECIMEN
MOVES IN ± Y-DIRECTION TO IMPACT AGAINST COUNTERFACE
v
FIBERS IN Y-DIRECTION
NORMAL APPROACH VELOCITY
FIBERS IN X-DIRECTION

Fig. 12



COTAC (Co-10Ni-12Ta-.75C)

Fig. 13

EFFECT OF EXTENDED CYCLES
COTAC (Co-10Ni-12Ta-.75C)
V = 4.2 ms^{-1}

MODE 1 Δ
MODE 2 o

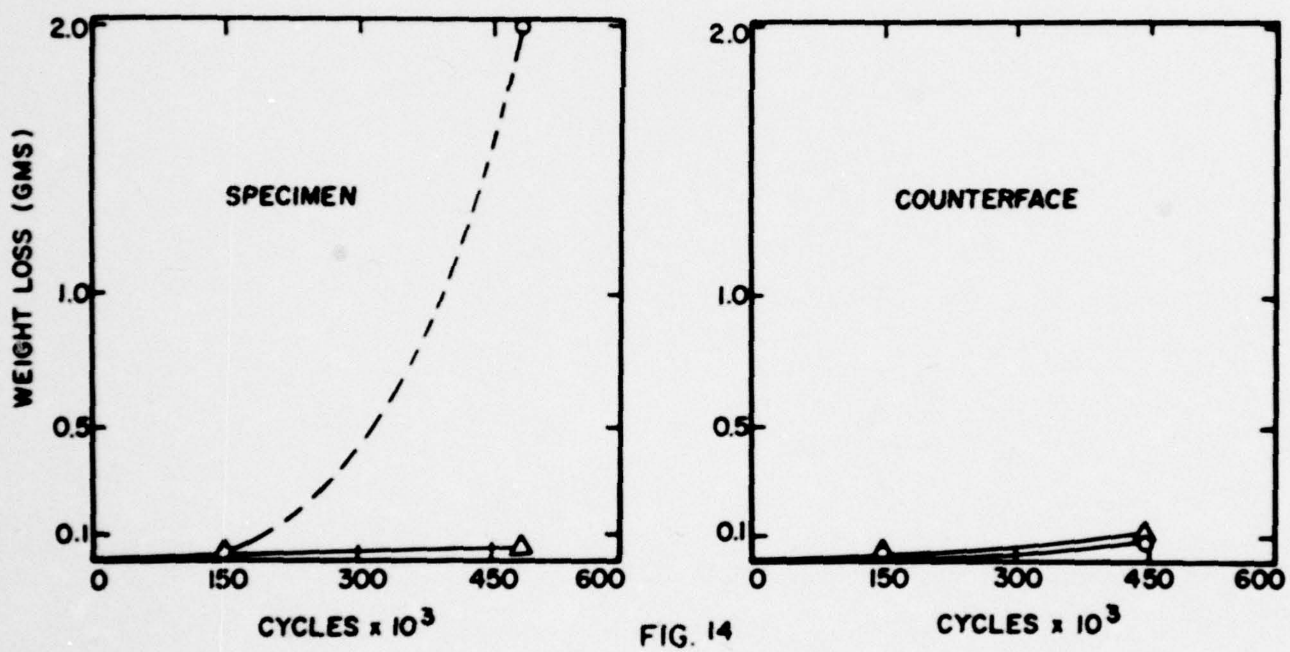
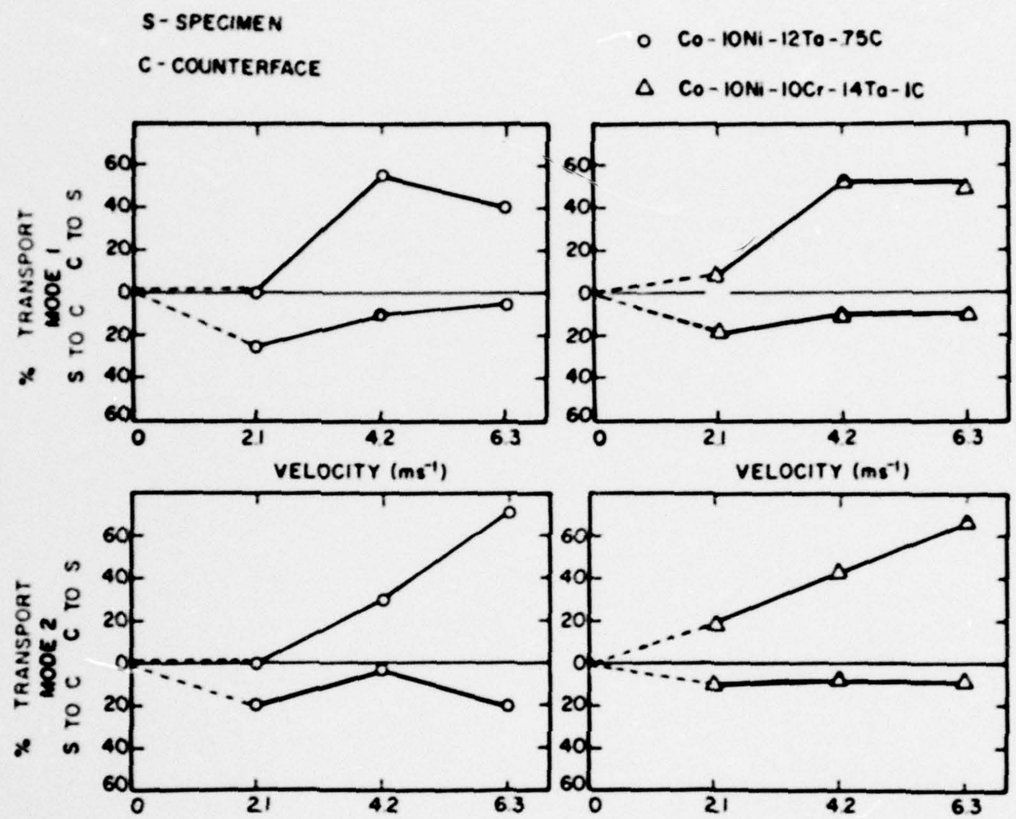


FIG. 14



MATERIAL TRANSPORT

Fig. 15

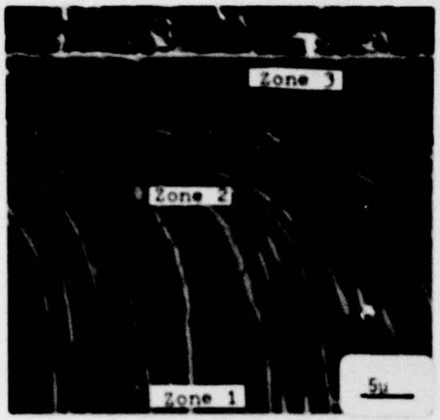


FIGURE 16. Micrograph (SEM), Co-10Ni-12Ta-.75C, sectioned after wear, (V=2.1 ms^{-1}, σ=18.6MPa, N=150,000 cycles), Mode 1, note the reorientation of broken up TaC fibers into the sliding direction.

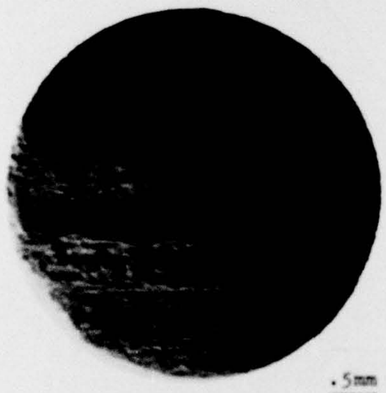


FIGURE 17. Micrograph (SEM), surface of Cu-0.81Cr specimen (V=2.1 ms^{-1}, σ=11.2MPa, N=250,000 cycles)

FIGURE 18. Micrograph (SEM), subsurface section of Cu-0.81Cr after wear (V=2.1 ms^{-1}, σ=11.2MPa, N=250,000 cycles), note disintegration of grains and strong orientation into sliding direction.

FIGURE 19 Micrograph (SEM), subsurface section of Cu-0.81Cr after wear (V=2.1 ms^{-1}, σ=11.2 MPa, N=250,000 cycles), note plastic formation of wear "lip".

FIGURE 20 Micrograph (SEM), subsurface section of Cu-0.81Cr after wear (V=2.1 ms^{-1}, σ=11.2 MPa, N=250,000 cycles), note discrete layering of specimen and 17-4 PH counterface materials.

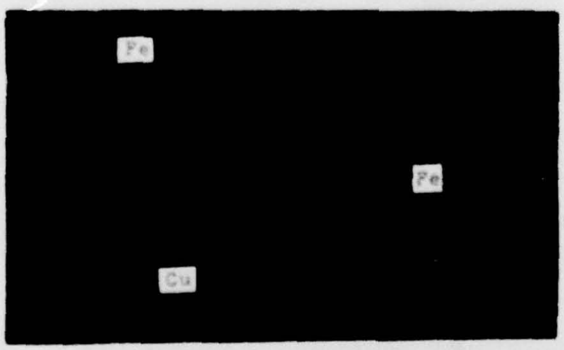
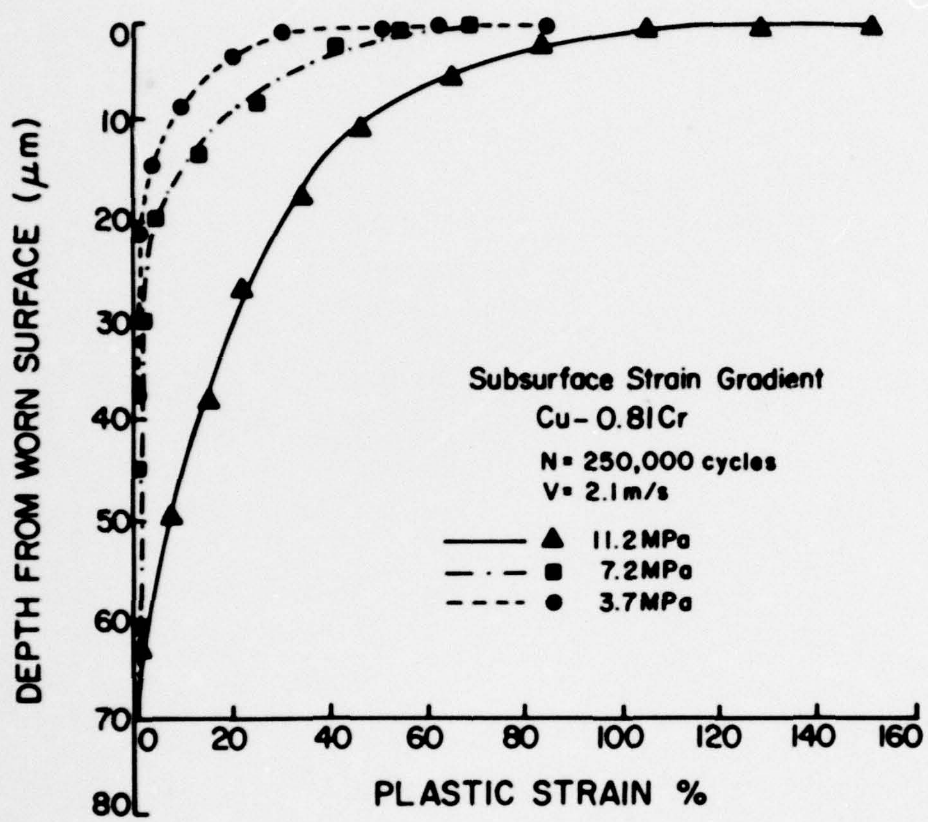


FIGURE 21 Element distribution map of FIGURE 20 as found by energy dispersive x-ray methods, note the highlighted near-surface bands of 17-4 PH material.



Subsurface Strain Gradient
Cu-0.81Cr
N= 250,000 cycles
V= 2.1m/s
11.2MPa
7.2MPa
3.7MPa
DEPTH FROM WORN SURFACE (μm)
0
10
20
30
40
50
60
70
80
0
20
40
60
80
100
120
140
160
PLASTIC STRAIN %

Fig. 22

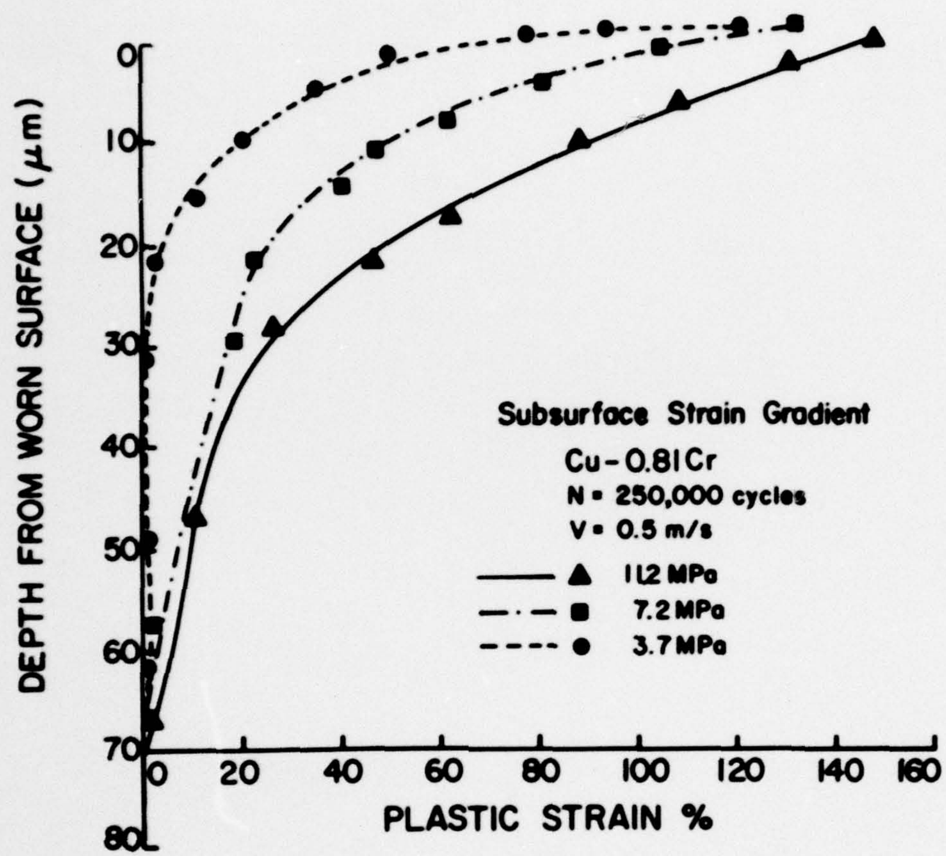

DEPTH FROM WORN SURFACE (μm)
0
10
20
30
40
50
60
70
80
0
20
40
60
80
100
120
140
160
PLASTIC STRAIN %
Subsurface Strain Gradient
Cu-0.81Cr
N = 250,000 cycles
V = 0.5 m/s
11.2 MPa
7.2 MPa
3.7 MPa

Fig. 23

INSTITUTE OF MATERIALS SCIENCE

The Institute of Materials Science was established at The University of Connecticut in 1966 in order to promote the various fields of materials science. To this end, the State of Connecticut appropriated $5,000,000 to set up new laboratory facilities, including approximately $2,150,000 for scientific equipment. In addition, an annual budget of several hundred thousand dollars is provided by the State Legislature to support faculty and graduate student salaries, supplies and commodities, and supporting facilities such as various shops, technicians, secretaries, etc.

IMS fosters interdisciplinary graduate programs on the Storrs campus and at present is supporting five such programs in Alloy Physics, Biomaterials, Crystal Science, Metallurgy, and Polymer Science. These programs are directed toward training graduate students while advancing the frontiers of our knowledge in technically important areas.